

AD-A063 654

LEHIGH UNIV BETHLEHEM PA MATERIALS RESEARCH CENTER
DISCUSSION OF 'A MODEL OF FATIGUE CRACK GROWTH IN POLYMERS'.(U)
DEC 78 R W HERTZBERG, M D SKIBO, J A MANSON N00014-77-C-0633

F/G 11/9

UNCLASSIFIED

TR-1

NL

1 OF 1
AD
AD 63 654

END
DATE
FILMED
3-79
DDC



AD A063654

LEVEL

OFFICE OF NAVAL RESEARCH

Contract N00014-77-C-0633

12

Task No. NR 356-670

9 TECHNICAL REPORT NO. 1

14 TR-1

6 Discussion of "A Model of Fatigue Crack Growth in Polymers"

by

R. W. Hertzberg, M. D. Skibo, John A. Manson, and J. K. Donald

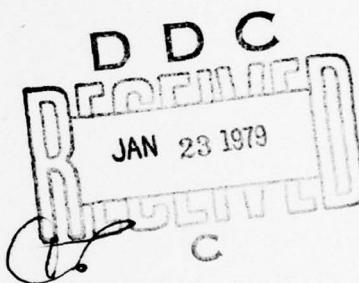
Prepared for Publication

in the

Journal of Materials Science

10

Richard W./Hertzberg,
Michael D./Skibo, John A./Mason
J. K./Donald



DDC FILE COPY

use → Materials Research Center
Lehigh University
Coxe Laboratory, #32 = *Location*
Bethlehem, PA 18015

11 28 December 28, 1978

12 16p

Reproduction in whole or in part is permitted for
any purpose of the United States Government

This document has been approved for public release
and sale; its distribution is unlimited

79 01 19 03 9
408 206 Sun

REPORT DOCUMENTATION PAGE		READ INSTRUCTIONS BEFORE COMPLETING FORM
1. REPORT NUMBER Technical Report No. 1	2. GOVT ACCESSION NO.	3. RECIPIENT'S CATALOG NUMBER
4. TITLE (and Subtitle) A Model of Fatigue Crack Growth in Polymers		5. TYPE OF REPORT & PERIOD COVERED Technical Report Interim
		6. PERFORMING ORG. REPORT NUMBER
7. AUTHOR(s) R. W. Hertzberg, M. D. Skibo, John A. Manson, and J. K. Donald		8. CONTRACT OR GRANT NUMBER(s) N00014-77-C-0633
9. PERFORMING ORGANIZATION NAME AND ADDRESS Materials Research Center Lehigh University, Coxe Lab., #32 Bethlehem, PA 18015		10. PROGRAM ELEMENT, PROJECT, TASK AREA & WORK UNIT NUMBERS NR 356-670
11. CONTROLLING OFFICE NAME AND ADDRESS Office of Naval Research (Code 472) Arlington, Virginia 22217		12. REPORT DATE December 28, 1978
		13. NUMBER OF PAGES 11
14. MONITORING AGENCY NAME & ADDRESS (if different from Controlling Office)		15. SECURITY CLASS. (of this report)
		16a. DECLASSIFICATION/DOWNGRADING SCHEDULE
16. DISTRIBUTION STATEMENT (of this Report) Approved for public release; distribution unlimited. Reproduction in whole or in part is permitted for any purpose of the United States Government.		
17. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, if different from Report)		
18. SUPPLEMENTARY NOTES		
19. KEY WORDS (Continue on reverse side if necessary and identify by block number) Fatigue Crack Propagation, Polymer		
20. ABSTRACT (Continue on reverse side if necessary and identify by block number) → A model was previously proposed by Hertzberg, Skibo, and Manson to account for the sensitivity of fatigue crack growth rates to frequency. The model was based on the hypothesis that the mechanical β transition controls the fatigue deformation and on the use of time-temperature relationships to determine the match between the β transition frequency and temperature and the frequency and temperature of the test. An alternate model was recently proposed by Williams, who interpreted the frequency sensitivity in terms of		

79 01 19 039

20. (Continued)

the sensitivity of fracture toughness (and Young's modulus) to frequency. In this communication, results of new experiments with nylon 66 are presented and discussed. Since Young's modulus was independent of test frequency for nylon 66, and for several other polymers, it was concluded that the authors' earlier hypothesis was not invalidated.

ACCESSION for

NTIS	White Section <input checked="" type="checkbox"/>
DDC	Buff Section <input type="checkbox"/>
UNANNOUNCED	<input type="checkbox"/>
JUSTIFICATION	<input type="checkbox"/>
BY	
DISTRIBUTION/AVAILABILITY COPIES	
DIR.	SP. CHL
P1	

DISCUSSION OF "A MODEL OF FATIGUE CRACK GROWTH IN POLYMERS"

by

Richard W. Hertzberg*

Michael D. Skibo**

John A. Manson*

J. K. Donald***

Williams has proposed an interesting model to describe fatigue crack propagation (FCP) in polymeric solids and to account for a number of experimental observations.¹ The purpose of this communication is to (1) examine the basic assumptions underlying the model, (2) compare recent data with values predicted from the model and (3) present alternate explanations for polymer fatigue behavior.

The first assumption is that upon unloading and reloading a craze at the crack tip, some of the craze ligaments become damaged, thereby reducing the craze stress σ_c . From this a two-stage craze zone is envisioned in which the newly formed craze material at the craze tip experiences a stress σ_c while the remaining part of the craze sustains a lower stress, $\alpha\sigma_c$. Use of this assumption leads to values of σ_c and $\alpha\sigma_c$ for several polymers in the ranges (325-720) MPa and (29-2,016) MPa, respectively (see Table II in References 1 and 2). In contrast, use of the Dugdale plastic strip formulation leads to values typically in the range of 40-80 MPa.² Since experimental values for crazing stresses are

* Materials Research Center, Lehigh University, Bethlehem, PA 18015

** Formerly with Materials Research Center, Lehigh University and currently Staff Scientist, Sandia Corporation, Livermore, CA 94550

***Del Research Corporation, Professional Services Group, Philadelphia Suburban Corporation, Hellertown, PA 18055

comparable for each of a number of polymers, it seems unlikely that the two-stage model can be generally valid, at least as currently stated.

Instead we suggest that the bulk of the craze experiences a uniform stress, σ_c ,³ similar to that postulated in the Dugdale plastic strip model.

We certainly agree with Williams that cyclic-stress-induced weakening will take place in some of the fibrils that span the craze. However, we postulate that the load across the craze will be redistributed among the remaining unbroken craze fibrils.³ These fibrils are then envisioned to stretch further, thereby leading to enhanced orientation hardening. With further cycling, additional fibrils are expected to break and the remaining ligaments would correspondingly become more highly oriented. We suggest further that a steady-state balance is struck between these two competitive processes--weakening through fibril fracture, and strengthening due to orientation hardening of the remaining fibrils--with consequent development of a constant stress, σ_c , across the craze. Quantitatively, this stress level should correspond to the product of load-bearing fibril strength σ_f and fibril volume fraction v_f so that $\sigma_c \approx \sigma_f v_f$. By way of confirmation, we find that computations involving fracture band widths, based on the Dugdale model, permit one to infer uniform craze stresses in several polymeric solids that, as mentioned above, are in good agreement with values reported in the literature by others.²

Proceeding further, we believe that the weight of evidence does not support Williams' explanation for the sensitivity of FCP rates to test frequency. He argues that polymer crack growth rates may be given by

$$\frac{da}{dn} = A \left(\frac{\Delta K}{K_c} \right)^n \quad (1)$$

where $\frac{da}{dn}$ = fatigue crack growth rate

A, n = material property

ΔK = stress intensity factor range

K_c = fracture toughness

Indeed relationships of this form has been proposed by Wnuk⁴ and supported by extensive experimental findings by several groups.⁵⁻⁷ Using Equation 1, Williams then proposes that the sensitivity of FCP rate to frequency is controlled by the strain-rate sensitivity of K_c , the latter being given by

$$K_c^2 = E \cdot \sigma_{ys} \cdot COD \quad (2)$$

where E = elastic modulus

σ_{ys} = yield strength

COD = crack opening displacement.

Earlier Williams⁸ assumed that the yield strain ϵ_y could be estimated from Hooke's Law as

$$\epsilon_y = \frac{\sigma_{ys}}{E} \quad (3)$$

so that $K_c = E/\text{COD} \cdot \epsilon_y$. The use of Eq. 3 in this situation seems questionable, especially since the modulus values used were defined at a strain of 3½ percent.

In any case, using yield strength and secant modulus data, along with Hooke's Law, Williams concluded that the yield strain was insensitive to strain rate and that the frequency sensitivity of K_c was due only to strain-rate-induced changes in E.

We disagree with this analysis in principle and on the basis of lack of correlation with both existing data and new test results reported below. First, by defining a secant modulus at a strain of 3½ percent and assuming

a true elastic limit at a much lower strain level, one would expect the secant modulus to be strongly sensitive to the yield strength (see Figure 1). It would not be surprising then to find the strong frequency sensitivity of E that was reported by Williams.⁸ On the other hand, moduli of typical glassy polymers are stated to be relatively insensitive to strain-rate,⁹ and even semi-crystalline polymers show relatively small time-dependent changes below T_g .¹⁰ If da/dN is to be changed by an order of magnitude (as is the case with some polymers¹¹⁻¹³) then E would have to change by a factor of 1.33 even if we assume the high value of 8 for the exponent n in Eq. 1 and assume that the frequency sensitivity of K_c is due only to strain-rate-induced changes in E.

Second, in order to examine directly the frequency dependence of E, we recently obtained compliance measurements from standard compact-tension samples, using the same geometry used to generate our FCP test results. These measurements were obtained under cyclic loading conditions at test frequencies ranging from 0.1 to 100 Hz. With the aid of data processing from an on-line PDP-8e computer, 20 to 100 individual data points (depending on test frequency) corresponding to specimen load P and associated crack opening displacement v were identified for each loading cycle. These values were used to establish a best-fit slope of the $\Delta P-\Delta v$ line. Between 2 and 40 such slopes were then used to define a final average slope. For a given crack length to specimen width ratio a/W and specimen thickness B, the modulus of elasticity of each sample could then be computed from the known compliance calibration relationship for the test specimen. It is of particular note that for most of the materials tested, the computed value of elastic modulus did not change to any significant degree (Table I). Since the body of the compact tension sample is predominantly elastic and

experiences very small strains, the results from these compliance measurements should reveal the material's elastic modulus rather than the secant modulus which Williams reported at a strain level of 3.5 percent. Because the FCP process and the associated crack tip stress intensity conditions are controlled by the elastic volume surrounding the small crack tip zone, the values of E reported here are considered to be more meaningful in assessing the FCP frequency dependence on $E(\dot{\epsilon})$. The reported E values are in some cases higher than those normally reported, based on conventional stress-strain data, but in general agreement with values reported based on dynamic mechanical data.¹⁴ Higher values would not be surprising since the specimen strains are very low.

The results confirm the relative insensitivity to frequency anticipated for diverse polymers. Note in poly(vinyl chloride) (PVC), polystyrene (PS), and poly(phenylene oxide) (PPO), that the measured values of E changed by only about one percent for each of several decade changes in cyclic test frequency. Compare this lack of modulus-frequency sensitivity with the previously documented strong FCP frequency sensitivity for these materials.¹¹⁻¹³ Clearly, frequency-induced changes in E cannot as a general rule account for the large frequency sensitivity factors (FSF) reported. Instead, we maintain our view that FCP frequency sensitivity is largely controlled by a resonance condition between test machine frequency and the frequency of molecular segmental motions associated with the β damping peak.¹¹⁻¹³

For the case of commercially prepared PMMA, we report an 8-10 percent change in E per decade change in test frequency which is considerably smaller than that reported by Williams, based on the 3.5 percent secant modulus.⁶ This highlights the difference in E based upon different strain level reference points. When the $E(\dot{\epsilon})$ values are included in Eq. 1 along

with the material parameter n , the computed change in FCP rate per decade change in frequency is less than that actually measured. In fact, a similar $E(\dot{\epsilon})$ sensitivity was found in laboratory-cast PMMA and, yet no sensitivity of FCP to frequency was found in this material (Table I). Finally, tests were conducted on dry nylon 66 to establish both the FSF and $E(\dot{\epsilon})$. We found no change in E as a function of test frequency and no sensitivity of FCP to frequency (see Figure 2), as was also the case for nylon 66 containing an unknown amount of moisture.¹¹ The reason for the difference in frequency sensitivity of FCP rates between our results and those of El-Hakeem¹⁵ (see reference 1, Figure 9) for dry nylon 66 is not clear at this time.

We certainly agree that the value of E to be used is the value at the test frequency selected. However while the secant modulus may show a strong strain rate sensitivity (presumably related to the strong strain-rate sensitivity of σ_{ys}), we conclude that the secant modulus is not the appropriate parameter for evaluating modulus effects on fatigue crack propagation behavior. Rather, we conclude that the modulus defined at small elastic strains is a more meaningful parameter for this purpose. However, the fact that $E(\dot{\epsilon})$ does not explain the FCP dependence on test frequency, supports our previously reported hypothesis that β -peak related segmental motions hold the key to the FCP frequency sensitivity in polymeric solids.¹¹⁻¹³

Acknowledgements

For the nylon 66 studies, the work was supported in part by the Office of Naval Research. Support of the PVC work by the National Science Foundation and of the earlier work by the Army Research Office is also acknowledged. Finally, we are indebted to one of the authors (J. K. McDonald) for providing invaluable assistance in conducting the compliance tests in his laboratory.

References

1. J. G. Williams, J. Mat'l. Sci., 12, 1977, p. 2525.
2. R. W. Hertzberg, M. D. Skibo and J. A. Manson, ibid, p. 531.
3. R. W. Hertzberg, M. D. Skibo and J. A. Manson, ASTM E-9 Symposium on Fatigue Mechanisms, May 1978, Kansas City, Missouri.
4. M. P. Wnuk, Prog. Rep., NASA Grant NGR 42-003-006, May 1971.
5. J. A. Manson and R. W. Hertzberg, CRC Crit. Rev. in Macromol. Sci., 1, 1973, p.
6. J. A. Manson, S. L. Kim, M. D. Skibo and R. W. Hertzberg, unpublished results, Lehigh University.
7. G. C. Martin and W. W. Gerberich, J. Mat. Sci., 11 (1976), 231.
8. J. G. Williams, Int. J. Fract. Mech., 8(4), 1972, p. 393.
9. R. N. Haward, The Physics of Glassy Polymers, R. N. Haward, Ed., Holstead Press, New York, 1973, pp. 41-43.
10. E. G. Bobalek and R. M. Evans, SPE Trans., 1, 1961, p. 93.
11. R. W. Hertzberg, J. A. Manson and M. D. Skibo, J. Polym. Eng. Sci., 15(4), 1975, p. 252.
12. R. W. Hertzberg, M. D. Skibo and J. A. Manson, Proc. 4th Int. Conf. Fracture, 3, Waterloo, Ontario, 1977, p. 1127.
13. R. W. Hertzberg, J. A. Manson and M. D. Skibo, Polymer, 19(3), 1978, p. 358.
14. N. G. McCrum, B. E. Read and G. Williams, Anelastic and Dielectric Effects in Polymeric Solids, John Wiley, New York, 1967.
15. H. M. El-Hakeem, Ph.D. Thesis, London University, 1975.

MATERIAL	0.1 Hz	1 Hz	10 Hz	100 Hz	n	$^*(E_{10Hz}/E_{1Hz})^n$	FSF **
Nylon 66 Dry	4160 MPa	4190 MPa	4190 MPa	4140 MPa	6.4	1	1
PVC ($\bar{M}_w = 1.4 \times 10^5$)	4590	4560	4600	4670	4.2	1.05	2.3
PVC ($\bar{M}_w = 2.3 \times 10^5$) + 6% DOP)	4290	4340	4410	4470	5.0	1.08	-
NORYL	3410	3490	3550	3580	4.9	1.09	2
ABS	3100	3170	3210	3210	3.7	1.05	1
PS	3930	3990	4120	4200	2.8	1.09	2.2
PMMA (laboratory cast $\bar{M}_w = 1.9 \times 10^5$)	5210	5740	6320	6980	9.3	2.44	1
PMMA (commercial $\bar{M}_w = 1.6 \times 10^6$)	3590	3960	4320	8.0	2.02	2.6	
PC	3250	3270	3320		4.2	1.06	1

* Ratio of modulii at frequencies of 1 and 10 Hz

** M. D. Skibo, R. W. Hertzberg and J. A. Manson, Fracture 1977, vol. 3, ICF 4, Waterloo, Canada, June 1977.

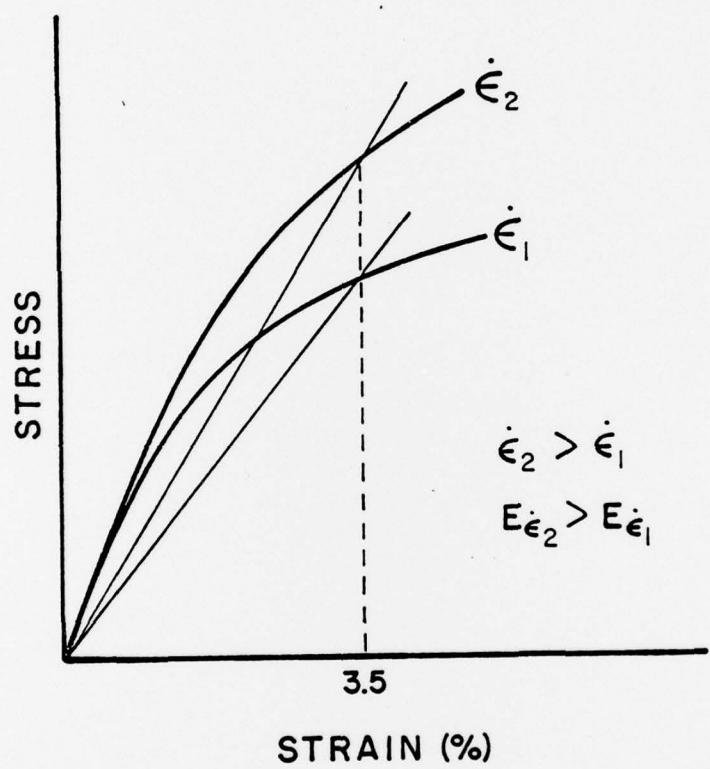


Figure 1

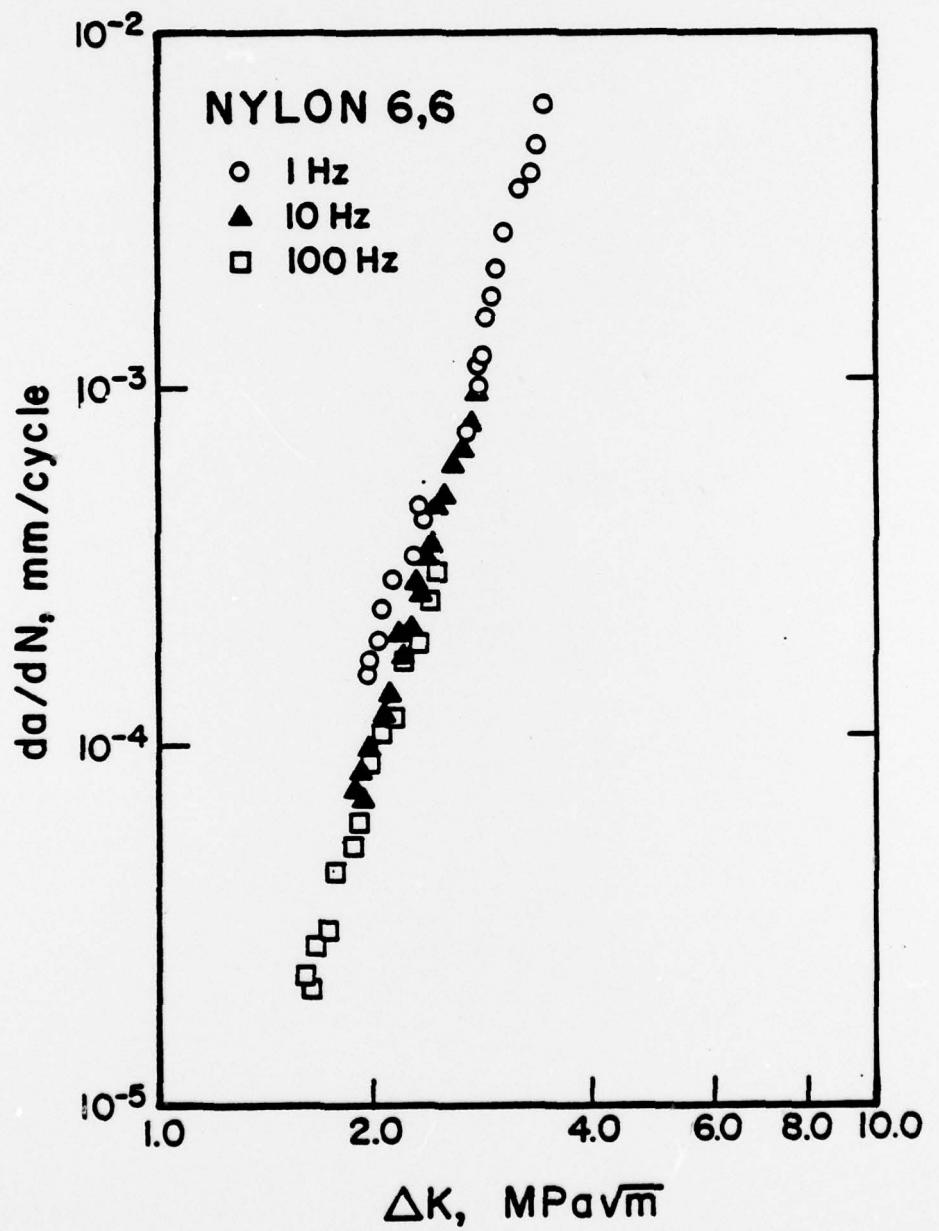


Figure 2

TECHNICAL REPORT DISTRIBUTION LIST, GEN

	<u>No.</u>	<u>Copies</u>		<u>No.</u>	<u>Copies</u>
Office of Naval Research 800 North Quincy Street Arlington, Virginia 22217 Attn: Code 472	2	Defense Documentation Center Building 5, Cameron Station Alexandria, Virginia 22314	12		
ONR Branch Office 536 S. Clark Street Chicago, Illinois 60605 Attn: Dr. George Sandoz	1	U.S. Army Research Office P.O. Box 1211 Research Triangle Park, N.C. 27709 Attn: CRD-AA-IP	1		
ONR Branch Office 715 Broadway New York, New York 10003 Attn: Scientific Dept.	1	Naval Ocean Systems Center San Diego, California 92152 Attn: Mr. Joe McCartney	1		
ONR Branch Office 1030 East Green Street Pasadena, California 91106 Attn: Dr. R. J. Marcus	1	Naval Weapons Center China Lake, California 93555 Attn: Dr. A. B. Amster Chemistry Division	1		
ONR Area Office One Hillidie Plaza, Suite 601 San Francisco, California 94102 Attn: Dr. P. A. Miller	1	Naval Civil Engineering Laboratory Port Hueneme, California 93401 Attn: Dr. R. W. Drisko	1		
ONR Branch Office Building 114, Section D 666 Summer Street Boston, Massachusetts 02210 Attn: Dr. L. H. Peebles	1	Professor K. E. Woehler Department of Physics & Chemistry Naval Postgraduate School Monterey, California 93940	1		
Director, Naval Research Laboratory Washington, D. C. 20390 Attn: Code 6100	1	Dr. A. L. Slafkosky Scientific Advisor Commandant of the Marine Corps (Code RD-1) Washington, D. C. 20380	1		
The Assistant Secretary of of the Navy (R, E&S) Department of the Navy Room 4E736, Pentagon Washington, D. C. 20350	1	Office of Naval Research 800 N. Quincy Street Arlington, Virginia 22217 Attn: Dr. Richard S. Miller	1		
Commander, Naval Air Systems Command Department of the Navy Washington, D. C. 20360 Attn: Code 310C (H. Rosenwasser)	1	Naval Ship Research and Development Center Annapolis, Maryland 21401 Attn: Dr. G. Bosmajian Applied Chemistry Division	1		
		Naval Ocean Systems Center San Diego, California 91232 Attn: Dr. S. Yamamoto, Marine Sciences Division	1		

TECHNICAL REPORT DISTRIBUTION LIST, 356A

<u>No.</u>	<u>Copies</u>	<u>No.</u>	<u>Copies</u>
Dr. Stephen H. Carr Department of Materials Science Northwestern University Evanston, Illinois 60201	1	Picatinny Arsenal SMUPA-FR-M-D Dover, New Jersey 07801 Attn: A. M. Anzalone Building 3401	1
Dr. M. Broadhurst Bulk Properties Section National Bureau of Standards U.S. Department of Commerce Washington, D. C. 20234	2	Dr. J. K. Gillham Princeton University Department of Chemistry Princeton, New Jersey 08540	1
Dr. T. A. Litovitz Department of Physics Catholic University of America Washington, D. C. 20017	1	Douglas Aircraft Co. 3855 Lakewood Boulevard Long Beach, California 90846 Attn: Technical Library C1 290/36-84 AUTO-Sutton	1
Dr. R. V. Subramanian Washington State University Department of Materials Science Pullman, Washington 99163	1	Dr. E. Baer Department of Macromolecular Science Case Western Reserve University Cleveland, Ohio 44106	
Dr. M. Shen Department of Chemical Engineering University of California Berkeley, California 94720	1	Dr. K. D. Pae Department of Mechanics and Materials Science Rutgers University New Brunswick, New Jersey 08903	1
Dr. V. Stannett Department of Chemical Engineering North Carolina State University Raleigh, North Carolina 27607	1	NASA-Lewis Research Center 21000 Brookpark Road Cleveland, Ohio 44135 Attn: Dr. T. T. Serofini, MS-49-1	1
Dr. D. R. Uhlmann Department of Metallurgy and Material Science Center for Materials Science and Engineering Massachusetts Institute of Technology Cambridge, Massachusetts 02139	1	Dr. Charles H. Sherman, Code TD 121 Naval Underwater Systems Center New London, Connecticut	1
Naval Surface Weapons Center White Oak Silver Springs, Maryland 20910 Attn: Dr. J. M. Augl Dr. B. Hartman	1	Dr. William Risen Department of Chemistry Brown University Providence, Rhode Island 02192	1
Dr. G. Goodman Globe Union Incorporated 5757 North Green Bay Avenue Milwaukee, Wisconsin 53201	1	Dr. Alan Gent Department of Physics University of Akron Akron, Ohio 44304	1

TECHNICAL REPORT DISTRIBUTION LIST, 356A

<u>No.</u>	<u>Copies</u>	<u>No.</u>	<u>Copies</u>
Mr. Robert W. Jones Advanced Projects Manager Hughes Aircraft Company Mail Station D 132 Culver City, California 90230	1	Dr. T. J. Reinhart, Jr., Chief Composite and Fibrous Materials Branch Nonmetallic Materials Division Department of the Air Force Air Force Materials Laboratory (AFSC) Wright-Patterson AFB, Ohio 45433	1
Dr. C. Giori IIT Research Institute 10 West 35 Street Chicago, Illinois 60616	1	Dr. J. Lando Department of Macromolecular Science Case Western Reserve University Cleveland, Ohio 44106	1
Dr. M. Litt Department of Macromolecular Science Case Western Reserve University Cleveland, Ohio 44106	1	Dr. J. White Chemical and Metallurgical Engineering University of Tennessee Knoxville, Tennessee 37916	1
Dr. R. S. Roe Department of Materials Science and Metallurgical Engineering University of Cincinnati Cincinnati, Ohio 45221	1	Dr. R. F. Helmreich Contract RD&E Dow Chemical Co. Midland, Michigan 48640	1
Dr. L. E. Smith U.S. Department of Commerce National Bureau of Standards Stability and Standards Washington, D.C. 20234	1	Dr. R. S. Porter University of Massachusetts Department of Polymer Science and Engineering Amherst, Massachusetts 01002	1
Dr. Robert E. Cohen Chemical Engineering Department Massachusetts Institute of Technology Cambridge, Massachusetts 02139	1	Professor Garth Wilkes Department of Chemical Engineering Virginia Polytechnic Institute and State University Blacksburg, Virginia 24061	1
Dr. David Roylance Department of Materials Science and Engineering Massachusetts Institute of Technology Cambridge, Massachusetts 02039	1	Dr. Kurt Baum Fluorochem Inc. 6233 North Irwindale Avenue Azusa, California 91702	1
Dr. T. P. Conlon, Jr., Code 3622 Sandia Laboratories Sandia Corporation Albuquerque, New Mexico 87115	1	Professor C. S. Paik Sung Department of Materials Sciences and Engineering Massachusetts Institute of Technology Cambridge, Massachusetts 02139	1
Dr. Martin Kaufmann, Head Materials Research Branch, Code 4542 Naval Weapons Center China Lake, California 93555	1		